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FRACTURE MECHANISMS IN ELECTRICAL INSULATOR PORCELAINS

by

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KEY WORDS: insulator strength, ceramic, porcelain cracks

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FRACTURE MECHANISMS IN ELECTRICAL INSULATOR PORCELAINS

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ABSTRACT

Fracture surface energy measurements for three electrical porcelain bodies indicated equivalent capabilities for supporting stress at crack tips in the glass phase. Superior strength of the alumina fortified porcelain was attributed to strain reduction afforded by the high elastic modulus of the system. The crack limiting ability of dispersed crystalline phases was assigned a secondary role in determining cross-bending strength. Apparent fracture surface energy of electrical porcelain was found to be almost twice that of soda lime glass.

INTRODUCTION

Much work has been done, on an empirical basis, to determine the mechanical properties of electrical insulator porcelains. Investigators such as Lundin[1] and Mattyasovszky-Zsolnay[2] have advanced several theories on the strength of porcelain, but it is evident that there is not a general consensus among researchers as to what strengthening or fracture mechanisms are present in a porcelain body. The character of a porcelain, with its 40 to 60 percent crystalline phases surrounded by a glassy matrix, is so complex as to be difficult to describe except on a macroscopic scale. Bulk properties are well known for porcelains, but there is still a large gray area encompassing the understanding of the fundamental mechanisms operating within the body.

The fracture surface energy (γ) of a brittle material is a fundamental property which is directly related to the ultimate tensile stress sustainable at the tip of a flaw in that material. It is defined according to fracture theory as the amount of energy required to form new surface by the extension of a preexisting flaw.

Complex materials such as polycrystalline or multiphase, glass-crystal ceramics exhibit non-planar fracture surfaces. Since the methods for determination of fracture surface energy do not take into account the fine scale irregularities of the fracture surface, it is appropriate to label the values determined for such materials as apparent or effective fracture surface energy (γ_e).

Fracture surface energies have been determined for various glasses[3], a number of single crystals[4], and a few polycrystalline ceramics[5]. Some work has been done to attempt to correlate the effective fracture surface energy in polycrystalline ceramics to the grain size in the ceramics.

The determination of effective fracture surface energies for several different electrical porcelain body compositions was undertaken with the hope that a better understanding of the mechanisms involved in fracture could be achieved.

THEORETICAL CONSIDERATIONS

The system used for the experiments was a modified form of the double cantilever beam described by Gilman[4] and Berry[6].

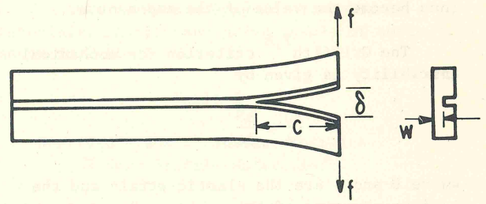

Fig. 1 Double Cantilever Sample

The basis for the experimental measurement is the fact that energy input into the sample will be stored as elastic strain energy in the two cantilever beam ends. If any excess of energy is stored in the two beam ends, over that amount of energy required to create new surface, the crack will propagate new surface until some equilibrium has been reached.

It is assumed that the relationship between the applied force (f), the deflection (δ), and the crack length (c) can be represented by the

equation

$$f = (ac^{-n})\delta$$

which is simply a general form of the equation

$$f = (3/2\ EIc^{-3})\delta$$

derived from classical beam theory. In the second equation E is the elastic modulus of the material and I is the moment of inertia of the cross-section of one half of the sample.

Since the uncracked region of the sample is not perfectly rigid, as would be described by a true cantilever beam system, and some energy is stored by elastic deformation of some portion of the beam beyond the crack tip, the value of the exponent n will be something less than the theoretical value of 3.

The empirical value of n for a particular system geometry may be derived from the expression

$$\log m = \log a - n \log c$$

where m is force divided by deflection and varies inversely as some power (n) of the crack length. The slope of a plot of log m versus log c will then become the value of the exponent n.

The Griffith[7,8] criterion for mechanical instability is given by

$$-\frac{\partial U}{\partial c} \geq \frac{\partial S}{\partial c}$$

where U and S are the elastic strain and the surface energies of the system. Based on this relationship, Berry[6] derived an expression for the calculation of fracture surface energy such that

$$\frac{f}{w} = \left(\frac{4\gamma_e}{n}\right)c$$

where w is the crack width. This was the equation used in the determination of fracture surface energies (γ_e) in the electrical porcelains tested.

EXPERIMENTAL TECHNIQUES

Three bodies were tested during the investigation. The first two bodies were conventional triaxial porcelains containing clay, feldspar and quartz. The most significant difference between the two triaxial bodies was that body "B" had specially sized quartz grains to avoid the presence of coarse grains evident in body "A". Body "C" contained alumina in place of the quartz. No quartz was added to this body per se. All three bodies were fired on the same schedule.

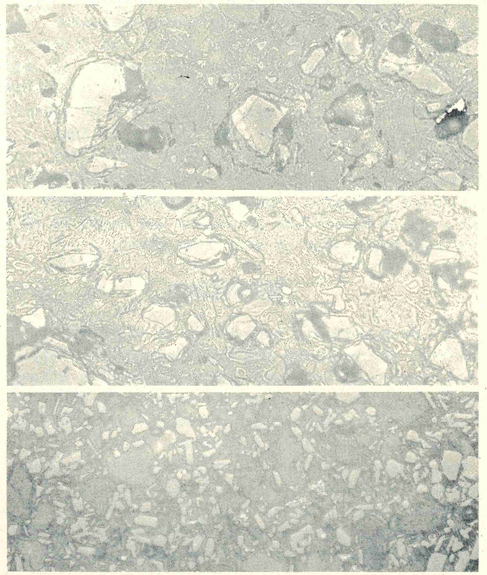

Fig. 2 Microstructural Comparison of Bodies A, B, and C (500X)

Porcelain slabs approximately 23 cm x 2.5 cm x 0.25 cm were cut from fired bars of all three compositions. A slot was diamond sawed through one half of the 0.25 cm dimension along the long axis of the slab. This slot was extended entirely through the 0.25 cm dimension for a distance of approximately 5 cm to allow a place for the crack to start. Loading holes were drilled in the ends of each beam arm.

The samples were then placed in an Instron tensile test machine and deflected at an 0.005 cm/min. rate until a change in the slope of the stress-strain curve indicated that a crack had been initiated. The deflection, measured by a strain gauge extensometer attached to the holding fixture, was halted and the force allowed to decay to an equilibrium value. Values for f and δ were taken from the x-y recorder attached to the test machine. A microscope, mounted on a cathetometer attached to the cross-head of the test machine parallel to the long axis of the sample, was used to determine the value of c.

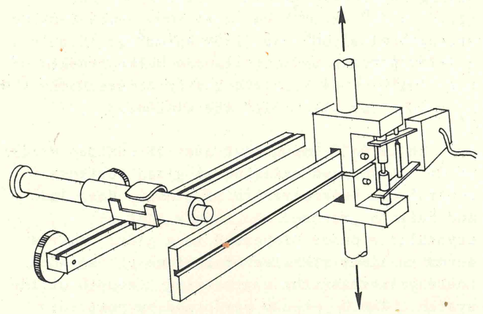

Fig. 3 Experimental Setup

Equilibrium generally took about 30 minutes to attain. After the equilibrium values for f, δ, and c were recorded, small increments of deflection (0.0005 cm) were added with a five minute wait between increments until the first force decay was noted. New values for f and δ were recorded at this fracture initiation point and the deflection was then increased at the 0.005 cm/min. rate and stopped when a mechanically unstable situation had been created, to allow the crack to propagate to a new equilibrium situation.

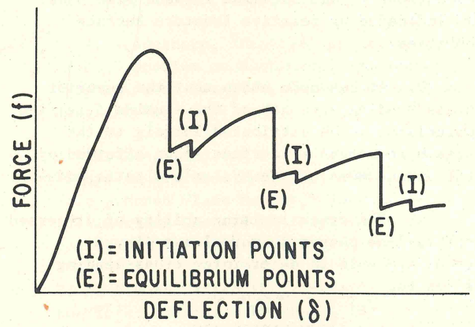

Fig. 4 Typical Stress Strain Curve for Double Cantilever Test

EXPERIMENTAL RESULTS

Fifty measurement points were used for each body. Results for bodies A and B showed an effective γ_e = 7000 erg/cm^2 and body C exhibited a γ_e = 7400 erg/cm^2. Further statistical analysis showed a significant overlap of the 90% confidence limits placed on the mean values indicating that the derived surface energies on all three bodies were not significantly different.

That portion of the experiment which dealt with fracture initiation showed a five percent average increase in the calculated stress at the crack tip over the value recorded at equilibrium. This calculation was based on the Griffith criteria for failure

$$\sigma_{crit.} = \left(\frac{2E\gamma}{\pi c}\right)^{\frac{1}{2}}$$

Ten measurements on each body showed good repeatability with the above mentioned fracture initiation values varying only between a three percent minimum and a six percent maximum.

DISCUSSION OF RESULTS

Previous investigations have yielded values of fracture surface energy for a number of materials. A representative sample of these is shown in Table I.

Table I

Observed Fracture Surface Energies of Some Brittle Materials[9]

Material and Grain Size	Fracture Surface Energy erg/cm^2
Fused SiO_2	4,320
Soda Lime Glass	3,820
Sapphire (1011)	3,000 - 8,000
MgO (7μ)	4,200
MgO (23μ)	16,000
MgO (130μ)	14,000
MgO (150μ)	7,900
Al_2O_3 (10μ)	18,000
Al_2O_3 (30μ)	27,000
Al_2O_3 (45μ)	46,000
Firebrick, Douglass X	30,000

Analysis of this table would indicate that glasses have inherently low fracture surface energies when compared to polycrystalline ceramics with little or no glass content. The three porcelain bodies used in this study

consisted of mullite and quartz (or alumina) crystals surrounded by a glassy matrix, which comprised 40 to 60 volume percent of the body. It is apparent the fracture path must intersect the continuous glass phase as crack propagation occurs. The fracture path may even develop preferentially in a manner such that it avoids the high modulus alumina particles as suggested by Hasselman and Fulrath[10].

Since the energy required to produce new surface is essentially the same for all three porcelain bodies, the stress sustainable at a crack tip is the same for each body. This is true at least to the extent that the surface area generated in each of the bodies is equivalent. An attempt was made to evaluate this factor by examining photomicrographic profiles of sections cut to intersect fracture surface for each body. Although a quantitative measure of roughness or surface area was not possible, no significant differences in roughness of profile were indicated.

With porcelain bodies having about the same ability to support stress concentration at crack tips, and with continuous glassy matrices available through which cracks must propagate, it is reasonable to suspect that the crystalline phases present act primarily to divert locally the path of propagating cracks. Furthermore, it appears that the bulk engineering strengths of the three bodies should be primarily dependent on the stress conditions imposed in the glassy phase.

When specimens are subjected to a cross-bending test the stress (S_T) developed in the outer fiber varies directly with its distance from the neutral fiber plane (y) and inversely as the deformed beam's radius of curvature (p) by the relationship[11]

$$S_T = (\frac{E}{p})y$$

where E is Young's modulus of elasticity. The addition of a material which has a large modulus of elasticity to a low modulus body would increase the modulus of the system. This would effectively strengthen the body by limiting the bulk strain (and hence, the stress) placed on the outer fiber, at a given load, as the dispersed high modulus material would absorb more of the applied stress per unit deformation.

The high Young's modulus of alumina (5.5×10^7 psi) serves to raise the effective modulus of the system when dispersed in a glass phase whose modulus is about 1.0×10^7 psi, while the addition of low modulus quartz particles does not. Quartz porcelains, having an elastic modulus of 1.0×10^7 psi (7.03×10^5 kg/cm^2) exhibit cross-bending strengths of approximately 10,000 psi (703 kg/cm^2) whereas the alumina porcelain, having an elastic modulus of 1.5×10^7 psi (1.054×10^6 kg/cm^2) has a minimum cross-bending strength of 15,000 psi (1054 kg/cm^2). This is a fifty percent increase in the bulk strength of the alumina body over the quartz bodies while the γ_e values for all bodies are similar.

It is well-known that size of surface cracks or Griffith flaws existing at glass surfaces control the bulk strength of glass. Hasselman and Fulrath proposed the theory that a crystalline phase dispersed in a glass matrix acted to limit the size of surface cracks and thereby increase the engineering strength of the system. The γ_e tests performed on porcelain bodies in this study were based directly on the Griffith criterion and showed only a five percent average increase in apparent stress bearing ability at the crack tip due to this flaw limiting effect.

CONCLUSIONS

1. The fracture surface energy measurements for three electrical porcelain bodies indicate they have approximately equivalent capabilities of supporting stress concentrations at crack tips. It is believed the continuous glass phase common to each body determines the common stress supporting level.

2. The energy required for propagation of already large cracks in these porcelains is almost twice that required in soda lime glass as indicated by relative fracture surface energies.

3. It has been shown that the superior cross-bending strength of the alumina fortified porcelain can be attributed largely to the strain reduction at surface flaws afforded by the enhancement of the system's elastic modulus.

4. The crack limiting ability of dispersed crystalline phases in porcelains play a secondary role in determining cross-bending strength.

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